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FATIGUE DAMAGE-STRENGTH RELATIONSHIPS IN COMPOSITE LAMINATES

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This report is Vol. I of a 2-part final report on the activities which were conducted under Contract No. F33615-81-K-3225, titled "Fatigue Damage-Strength Relationships in Composite Laminates" monitored by Dr. George Sendeckyj, Air Force Flight Dynamics Lab, Wright-Patterson Air Force Base, OH. The general objective of this program was to investigate the precise nature of the damage events caused by cyclic (fatigue) loading of composite laminates that are directly related to fracture of those laminates, to develop a

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conceptual understanding of how those damage events reduce the residual strength of the laminates, and to determine how the collective damage condition following long-term cyclic loading precipitates the final fracture event. The program was conducted in four phases: 1) the development of fatigue damage during long-term cyclic loading was investigated in such a way that the damage events which occur quite late in the fatigue life of the material, just prior to fracture, were identified and characterized; 2) the special contribution and nature of fiber fracture was investigated; 3) the nature of the collective damage condition which precipitates the fracture event under long-term fatigue loading was investigated; 4) a philosophy of fracture in composite laminates following severe damage development due to cyclic loading was developed. A wide variety of destructive and nondestructive experimental evaluation schemes were used in the program, and several analytical efforts were conducted to support the experimental activities.

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PREFACE

The work reported herein was performed under contract F33615-81-K-3225, Project 2307, Work Unit 2307N117, sponsored by the Flight Dynamics Laboratory of the Air Force Wright Aeronautical Laboratories, Wright-Patterson Air Force Base, Ohio 45433. Dr. G. P. Sendeckyj, AFWAL/FIBE, was the Air Force Program Monitor.

This is Volume I of the final report of work completed under the subject contract. Both Volumes I and II are summary volumes. Details of the results and related discussions can be found in an earlier report (AFWAL-TR-82-3103) and in the eleven papers published during the two year period of performance of the investigation. (See References 1-11.)

The authors express their sincere appreciation to the U.S. Air Force for their support of this investigation, to Dr. G. P. Sendeckyj for his technical guidance and support, to Bob Davis and G. K. McCauley in the Virginia Tech shop for their assistance in preparing specimens, equipment and photographs, and to Barbara Wengert for typing the manuscript.

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INTRODUCTION--THE PROBLEM

The study of composite laminate response under cyclic loading has been the objective of many distinguished investigators over a period of more than fifteen years. Detailed accounts of the damage that develops under such loading have been reported and presented in numerous papers and books. Two collected sources of such information are a recent ASTM Special Technical Publication and a review article by Stinchcomb, et al. (Refs. 13 and 21). These efforts have provided an excellent array of information regarding damage mechanisms and, to a lesser extent, residual properties after cyclic loading. Motivation for the present study comes from the fact that prior investigations have not specifically addressed the fatigue damage-strength relationships in composite laminates subjected to cyclic tensile loading amplitudes which corespond to lives of the order of one million cycles or more, what we will call the long-term fatigue behavior of laminates. Engineering applications commonly require that composite materials and components perform for long periods of time under fairly low cyclic load levels, and that an assessment of integrity of the material (or structure) be used to anticipate the residual strength and life of the component. Hence, the present investigation addressed the question of long-term behavior by attempting to answer two basic questions. First, since the residual strength is reduced by large amounts prior to failure at low load levels, exactly what are the mechanisms by which the strength is reduced and how do those mechanisms develop as a function of the life of the component? And second, what actually causes fracture or laminate failure in such a situation, i.e., how do the damage mechanisms combine to cause the final failure event?

The present investigation was conducted on composite laminate specimens made from high-modulus fibers and polymer matrix materials. A variety of laminate stacking sequences were investigated including $[0,90]_s$, $[0,\pm45]_s$, and several quasi-isotropic laminates, as well as other specialized arrangements. The central object of our attention can be best demonstrated by a brief discussion of Fig. 1. That figure shows a schematic representation of a classical S-N curve with residual strength variations superposed for a tensile load amplitude of 50% of the static ultimate strength. For purposes of discussion it has been assumed that the life of the specimen under test is one million cycles at that stress amplitude. In general, it can be assumed that long-term behavior will correspond to stress amplitude ratios that range between 40% and 70%. Although one tends to think about the low load amplitudes associated with long-term behavior as loads that do not cause very much damage allowing the specimens to "live longer", in fact, the damage associated with eventual failure at those load levels is greater than is experienced for higher load levels in the sense that residual strength and stiffness values are reduced to lower values before failure. In fact, the greatest possible damage corresponds to the longest possible <u>finite</u> life. It is exactly this fact that forms the basic theme for the present investigation. The object of the experimental part of our investigation was to capture, isolate, record, and understand the damage states which correspond to strength reductions of the order of 30 to 50%. The difficulty in conducting such tests is obvious. In order to obtain the data, one must continue to apply cyclic loads to the specimen

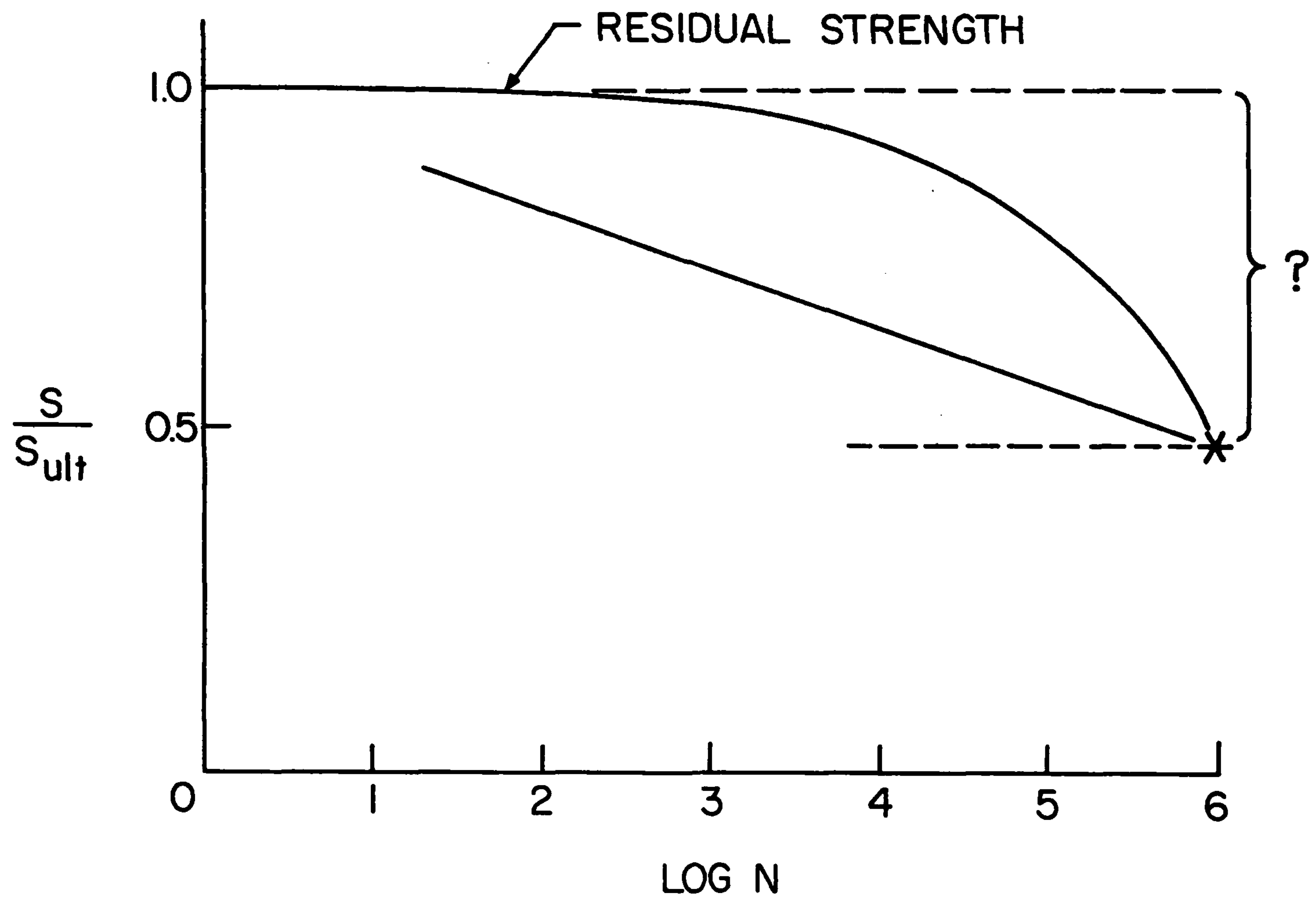

Figure 1. Schematic diagram of long-term fatigue strength reduction in composite laminates under fatigue loading.

until the specimen strength has been reduced to a level only slightly higher than the applied stress level. If too many cycles are applied, the specimen fails, destroying the data of interest. If too few cycles are applied, only a small amount of strength reduction is produced and it is not possible to investigate the damage state that immediately precedes the fracture event. Finally, since it is not possible to determine exactly how large the strength reduction is without actually failing the specimen during the test, one must depend upon a damage parameter to monitor the development of fatigue damage and to stop the test at the last possible moment for observation. The best method that the present investigators were able to find for monitoring damage development in this way was the continuous real-time measurement and observation of specimen stiffness. Using computer assisted testing techniques, stiffness changes provided a reproducible, quantitative, and easily interpretable indicator of damage development in the composite laminates tested. In this way, we were able to record (possibly for the first time) a variety of data which define the precise nature of the damage states that occurred immediately prior to fracture under long-term fatigue loading conditions.

This report (Vol. I) is an executive summary of a variety of investigative results which have been described and recorded in the literature (See Refs. 1-10). In the paragraphs that follow it is only intended to provide an integrated summary of those results, to highlight some of the salient aspects of those findings, and to provide some conclusions that summarize that part of the activity. A great many details will not be covered in this report. Interested readers are referred to the referenced work for that information.

MATRIX DAMAGE

The most widely characterized and discussed mode of damage due to fatigue loading of composite angle ply laminates is matrix cracking. The most easily identified and discussed type of matrix cracking consists of the formation of cracks transverse to the load axis which extend through the thickness of a given ply in a plane which is parallel to the fiber direction in the off-axis plies. This type of cracking is common and pervasive even at the stress levels associated with long-term behavior. For constant amplitude tension-tension loading, one might characterize the development of damage and the reduction of residual strength as shown in Fig. 2. As suggested by the figure, there is an early "stage of adjustment" to cyclic loading which is characterized by a rapid (and rapidly decreasing) rate of damage development--mostly transverse matrix crack formation. This type of transverse crack formation has received a great deal of attention and is, by comparison to other micro-events, fairly well described and understood. Formation of the cracks can be reasonably well anticipated by laminate analysis coupled with a common "failure theory" such as the maximum strain, Tsai-Wu or Tsai-Hill concepts. It was discovered in our Materials Response Laboratories some ten years ago that these matrix cracks form a regular array in each of the off-axis plies as quasi-static or cyclic loading continues. This array is a laminate property, i.e., it is completely defined by the properties of the individual plies, their thickness, and the stacking sequence of the variously oriented plies. These patterns are independent of loading history and other extensive variables and can be predicted by several types of analysis. We have named these crack

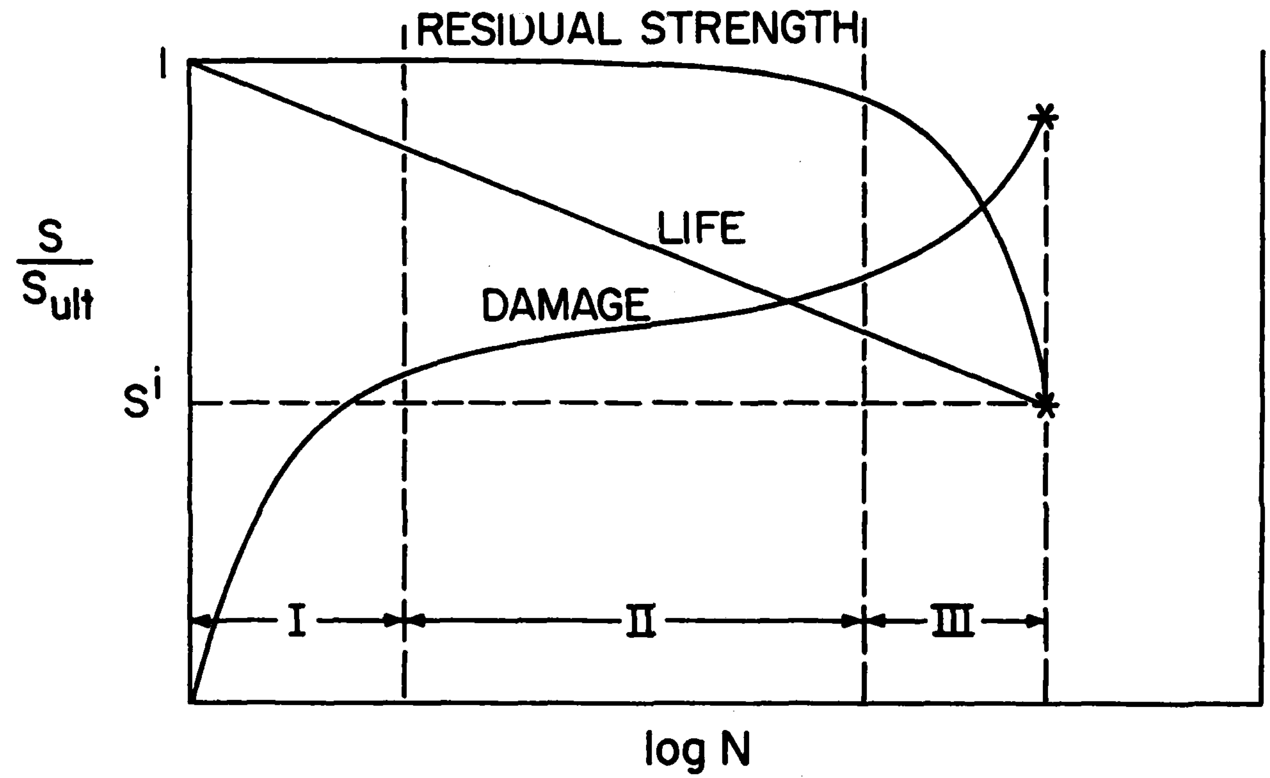

Figure 2. Schematic diagram of residual strength reduction and damage development in angle-ply composite laminates.

patterns "characteristic damage states" (CDS) for matrix cracking in laminates having off-axis plies subjected to load histories including tensile load excursions. A more detailed discussion of the CDS can be found in Refs. 13-17. The CDS is a stable damage state, and its formation is the reason for the sudden decrease in damage rate between regions 1 and 2 in Fig. 2.

The matrix cracks that formed the CDS introduce a reduction in the stiffness of the laminate since the cracked laminae carry less load than they did prior to cracking. For graphite epoxy quasi-isotropic laminates, that laminate stiffness change is of the order of 10% or somewhat less, while a comparable glass epoxy laminate stiffness change would be somewhat greater. These changes are generally not of large engineering consequence except as they affect vibration frequencies. Also, it should be noted that the transverse cracks, per se, do not reduce the residual strength of the laminate during fatigue loading since those cracks would form under any quasi-static loading history that would be used to define the static ultimate strength of the laminate. However, the stress redistributions which accompany the matrix cracks are, indeed, consequential. For quasi-static loading, this has been demonstrated by some fifteen years of experience throughout the composites community which indicates that a reasonably accurate engineering calculation of the strength of angle ply laminates can be made using laminate analysis and a computational scheme called the "discount method". In this scheme, the in-plane shear stiffness and the Young's modulus in the direction perpendicular to the fibers in off-axis plies in which matrix cracks form is reduced to zero (or a very small value) in a laminate analysis. The stress carried by those plies in the directions corresponding to the reduced stiffnesses is redistributed into the remaining unbroken plies (usually the zero degree ply or plies). The redistributed stress state calculated in this way is then used to determine the strength of a laminate which ultimately depends on the increased stress level in the last plies to fail and the strength of those plies (Ref. 1). Such a calculation actually represents the net section stresses calculated at a cross-section of the specimen coincident with the matrix cracks in the off-axis plies, since the stresses in the remaining unbroken plies (generally the zero degree plies) are nonuniform and display concentrations in the regions of the matrix cracks. The first demonstration (to our knowledge) of this fact was provided by Highsmith and Jamison who constructed a specialized high resolution moiré diffraction device designed by Professor Post at Virginia Tech and measured the strain distributions in the zero degree ply of several different laminates in regions near cracks in adjacent off-axis plies during quasi-static loading (6). An example of their results is shown in Fig. 3. That figure was produced by the interference between the reference beam and a beam which was incident on a diffraction grating having about 800 lines per millimeter which was bonded to the specimen surface. The cracks in the off-axis 90° plies of the $[0,90_3]_s$ glass epoxy specimen can be seen as white horizontal bars having a spacing of about 4 mm in the original photograph. The constant displacement diffraction lines are more dense in the region of the off-axis cracks, indicating a strain concentration in the zero degree ply which is being observed. The strain distribution in the neighborhood of off-axis cracks was measured in zero degree plies using this method and compared to several types of analytical results (Refs. 6 and 7). Since

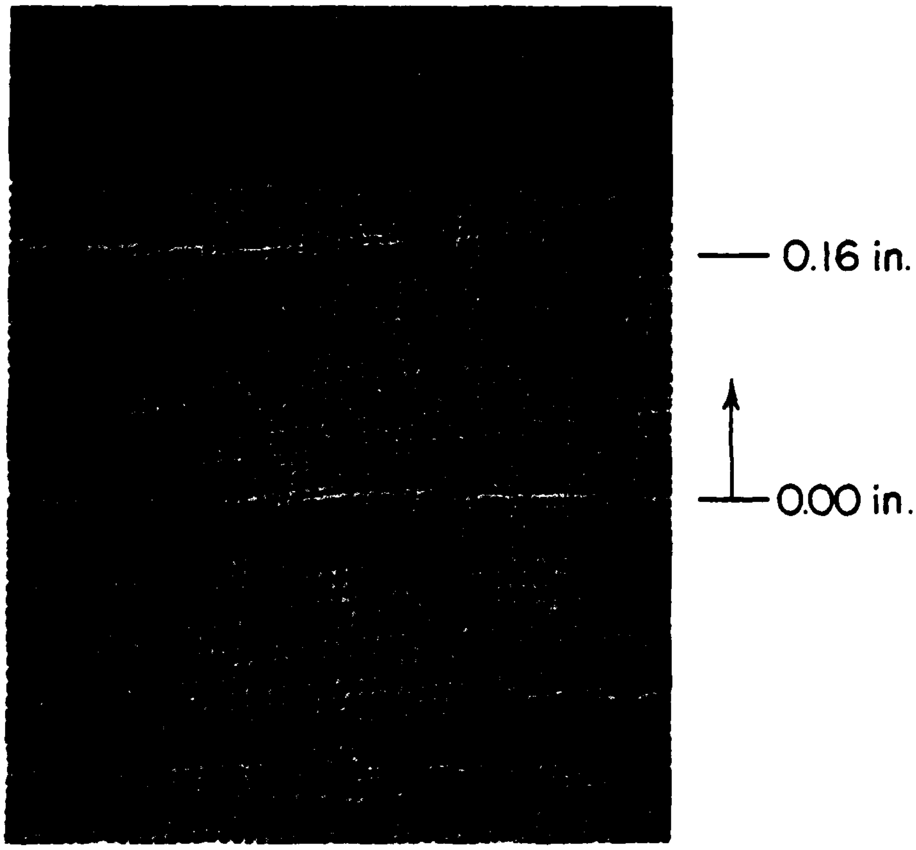

Fig. 3. Moiré fringe pattern of a $[0,90_3]_s$ laminate between cracked sections in the 90° plies.

the diffraction grating bonded to the zero degree ply measures the response of that ply averaged through its thickness, it is not surprising that those results correspond to the average net section stress computations obtained from laminate analyses or shear lag calculations. To that extent, then, some fifteen years of data appear to show that the net section strength of the zero degree plies in the neighborhood of off-axis ply cracks controls the quasi-static laminate strength, at least to an engineering approximation level.

It was once thought that these transverse matrix cracks were of no consequence in the process of residual strength reduction during cyclic loading. One of the most important findings of the present investigation is the discovery that such matrix cracks are the locations where subsequent severe damage localization and intensification occurs during cyclic loading and the further discovery that the local stress states created by such matrix cracks and subsequent damage development play a major role in strength reduction. As we will see in a later section, fiber fractures occur preferentially--indeed almost exclusively in the increased stress field around these matrix cracks, and in some cases internal delaminations develop in these regions driven by the interlaminar stresses that are induced by the matrix crack geometry (Refs. 1-10). Both of these phenomena will be discussed in a later section.

Another particularly interesting and significant finding of the present investigation, in addition to the nature of the influence of the primary matrix cracks discussed above, was the discovery of a secondary crack development process which appeared to various degrees in all laminates examined. These secondary cracks appear along the crack front of primary cracks, in directions which are perpendicular to the primary crack front. In the case of a $[0,90]_s$ laminate for example the secondary cracks would appear at the tip of the cracks in the 90° plies, initiating at the boundary between the zero degree plies and the 90° plies, and growing for short distances along the fiber direction in the zero degree plies as suggested by the diagram in Fig. 4. Two distinctive situations regarding the extent of the growth and formation of the secondary cracks appear to be evident. In one case, the global stress in the ply in which the secondary cracks are growing in the crack opening direction for those secondary cracks is tensile, and adds to the tensile stress created by the primary cracks. Such a situation is most easily demonstrated in the case of the $[0,90_2]_s$ laminate. Since the transverse normal stresses in the zero degree ply of that laminate under tensile loading are tensile, the secondary cracks which form in the location of the primary cracks in the 90° plies propagate quickly through the thickness of the zero degree plies and along the length. The result of such a process is shown in Fig. 5. That figure shows an enlargement of a radiograph of a fatigue damaged $[0,90_2]_s$ graphite epoxy laminate in which the primary matrix cracks (horizontal lines) and secondary matrix cracks (vertical lines) are clearly visible. Many of the secondary matrix cracks grow entirely through the thickness of the zero degree plies and essentially along the entire length of the specimen. The second distinctive case is created by the situation in which the global stress in the plies in which the secondary cracks form in the crack opening direction of the secondary cracks is compressive, and therefore opposes the tensile stress created by the primary matrix cracks. This is frequently the case when 45° plies are present.

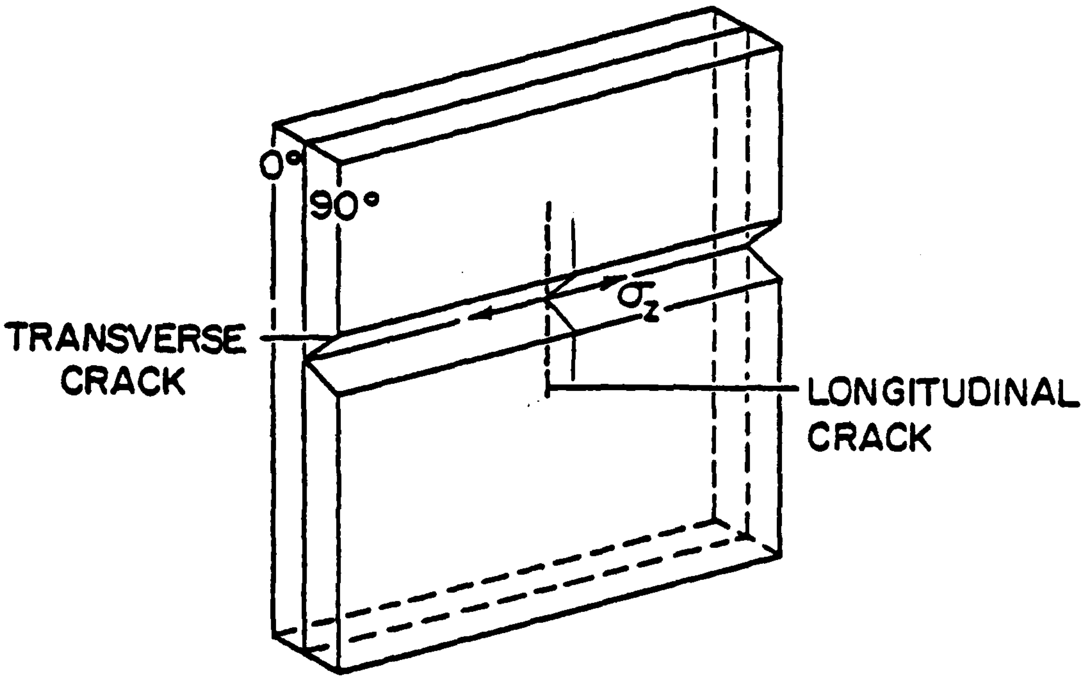

Figure 4. Zero Degree Stresses in the Neighborhood of a Transverse Crack.

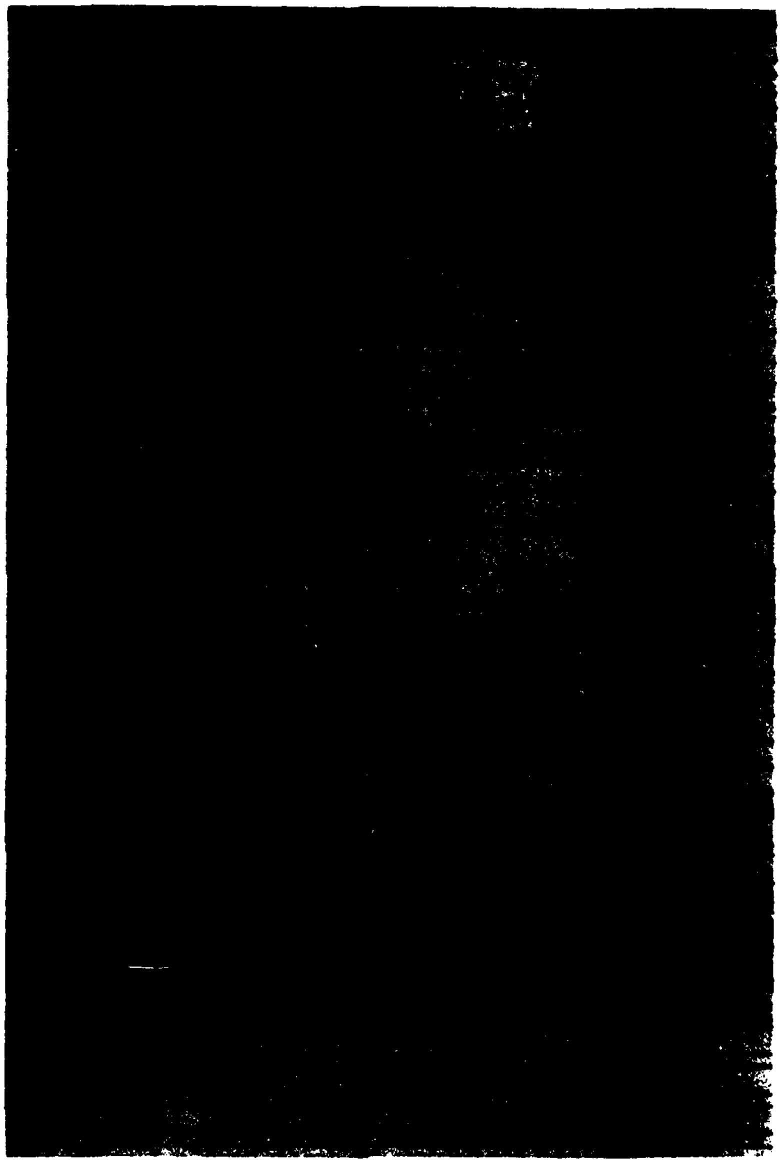

Figure 5. Radiograph showing primary (horizontal) and secondary (vertical) matrix cracks in a $[0,90_2]_s$ graphite epoxy laminate.

Figure 6 shows an example of secondary crack formation for such an instance. While the density of secondary cracks in this situation is very great, the extent of the cracks is small, quite closely restricted in every direction to the immediate vicinity of the primary crack front. In some cases, these secondary cracks form so close to one another that they nearly become a contiguous delaminated region. More will be said of this possibility later. Generally speaking, these secondary cracks initiate and grow quite late in the life of a long-term fatigue specimen, frequently in the last one-third or so of the total life. The present investigators did not have very much opportunity to characterize the secondary crack process or to establish its direct or indirect role in the reduction of residual strength or the final failure process. The investigators were struck, however, by the nearly omnipresent nature of the secondary cracks in graphite epoxy laminates in every stacking sequence examined (Refs. 1-4). Perhaps the greatest dilemma presented by the discovery of these secondary cracks is the question of how to determine the combined stress state created by the cross-crack pattern formed by the primary and secondary cracks at the interface between their respective plies. The determination of that stress field requires the solution of a complex three-dimensional boundary value problem at the interface between two orthotropic layers. Such an analysis has not yet been constructed. As we will see in subsequent sections, the position where the primary and secondary cracks cross is the location where severe subsequent damage localization occurs. Because of the generality of its occurrence, the present investigators believe that the cross-crack problem is generic in nature and should be the object of further investigation.

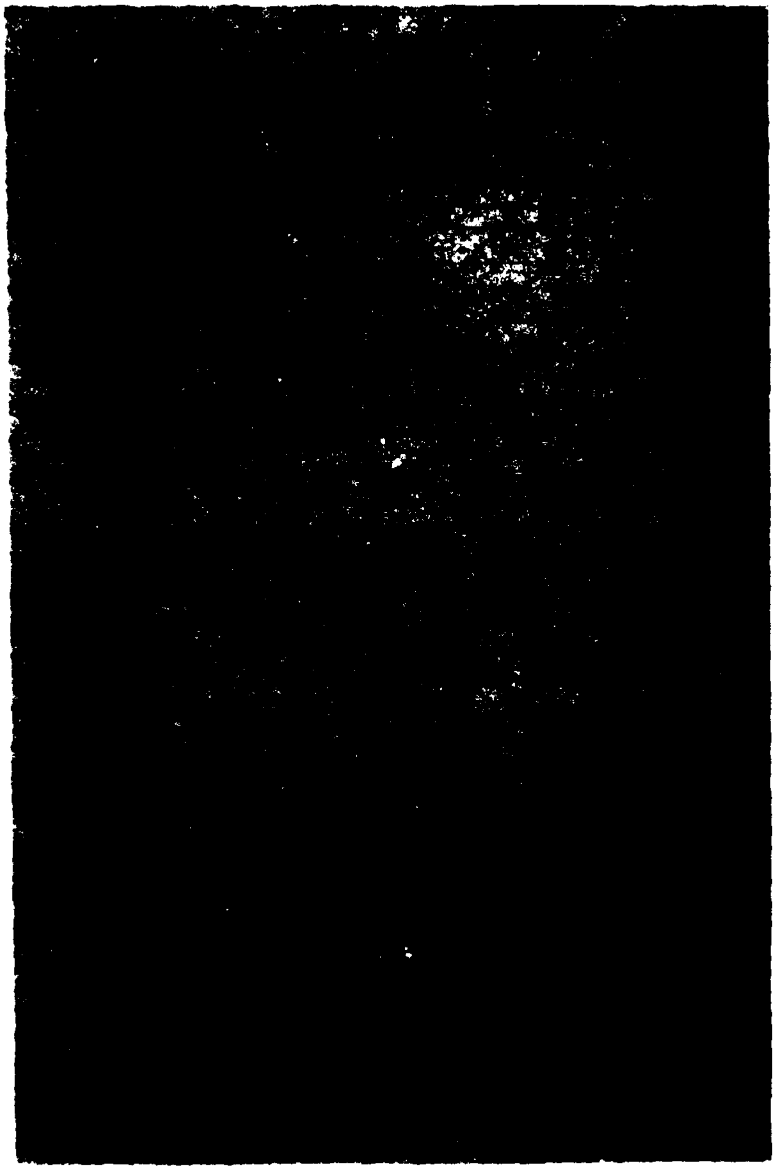

Figure 6. Radiograph showing a high density of secondary crack formation in a $[0,90,\pm 45]_s$ laminate.

FIBER DAMAGE

One of the damage modes least well characterized in the literature is that of fiber failure under cyclic loading. Some investigators have attempted to look at this important aspect of the problem in unidirectional materials under cyclic loading, but there is a nearly total dearth of information about fiber fracture in angle ply laminates under fatigue loading. One of the most important findings of the present investigation, in the opinion of the investigators, is the discovery that fiber fractures in angle ply laminates occur in a manner that is uniquely peculiar to a specific laminate and to the situation wherein matrix cracks in one ply intersect fibers in an adjacent ply at angles which are not parallel to those fiber directions. The salient aspects of the fiber fracture investigation are outlined below.

While a number of fiber fracture investigation and characterization schemes were used, the present study of fiber fracture was divided into two generic categories. One category dealt with fiber fractures that were observed in the interior of angle ply laminates and the other category dealt with fiber fractures that were observed to develop near the edges of specimens. The general nature and consequence of these two types of fiber fracture patterns was somewhat different.

From the standpoint of load carrying capability, most engineering components designed using composite materials are constructed in such a way that the fibers dominate the load carrying capability of the components. Because of this fact, a composite component will generally not rupture until a substantial number of fibers have been broken. Exceptions to this include instances where delamination can completely separate a composite laminate during service, but even then the final fracture event will probably involve a large amount of fiber fracture. The easiest place to observe such fractures is near a surface or edge of a component or laminate. In the present investigation, the first indications that fiber fractures were related to matrix cracks in some instances was provided by studies of broken fibers that could be observed at the edges of our coupon specimens. These specimens had ply orientations of $[0,\pm45,0_2,\pm45,0]_s$ and $[0,90,0,90_2,0,90,0]_s$ and were constructed from T300/914C graphite epoxy material. These specimens were supplied by the DFVLR, Institute for Structural Mechanics in Braunschweig, West Germany and the investigation was conducted by Dr. K. Schulte from that Institute who was conducting post-doctoral studies in the Materials Response Laboratory at Virginia Tech. The results of that investigation and some related work appear in Ref. 3.

The cyclic load amplitudes applied to these specimens was typically 70% of the static fracture stress in tension. Different load levels were investigated, and the extent of damage development was influenced by load level as one might expect, but the basic nature of the damage development patterns was not sensitive to the load level for amplitude ratios between about 70% and 85%. At these load levels, as discussed earlier, matrix cracks in off-axis plies develop quite rapidly and quickly saturate to form a stable regularly spaced pattern which we have called the characteristic damage state (CDS) which is characteristic of the laminate being tested. As the formation of the CDS is being completed, the fiber fracture process at the edge of the specimen began. While additional fiber fractures developed throughout the remainder of the test up until the final fracture event, a significant number of fiber fractures developed into a well defined pattern in the early stages of the life of the specimen, generally within the first third of the life. Examples of this are shown in Figs. 7a and 7b. In Fig. 7a, it can be seen that the cracks in the off-axis plies have formed a relatively regular spacing which has become nearly saturated to form the CDS as indicated earlier. It is also seen that a number of fiber fractures have developed in regions which appear to be associated with these matrix cracks. Figure 7b provides a more clear indication of that association. The fiber failures appear to be aligned with one another in groups which seem to be located at positions which are related to cracks in the matrix of the off-axis plies on either side of the zero degree ply. While the surface replication scheme cannot provide any evidence of a direct link between the fiber fractures and the matrix cracks, a substantial amount of data was developed during the present investigation which indicates that the fiber fracture pattern is related to the stress field developed by the matrix cracks in the off-axis plies.

As we mentioned earlier, this fiber fracture process continues throughout the remainder of the specimen test. However, the essential features are established in the early life of the specimen or component, and the fiber fracture pattern does not appear to lead to a localization of damage that eventually becomes critical in the process of fracture initiation. Instead, it appears that the fiber fractures tend to relax the stress concentrations associated with the notch effect caused by the matrix cracks in that region. The fiber fractures seem to be limited to a very small volume of material corresponding to a thin layer of one fiber thickness or so next to the surface of the specimen on the edge

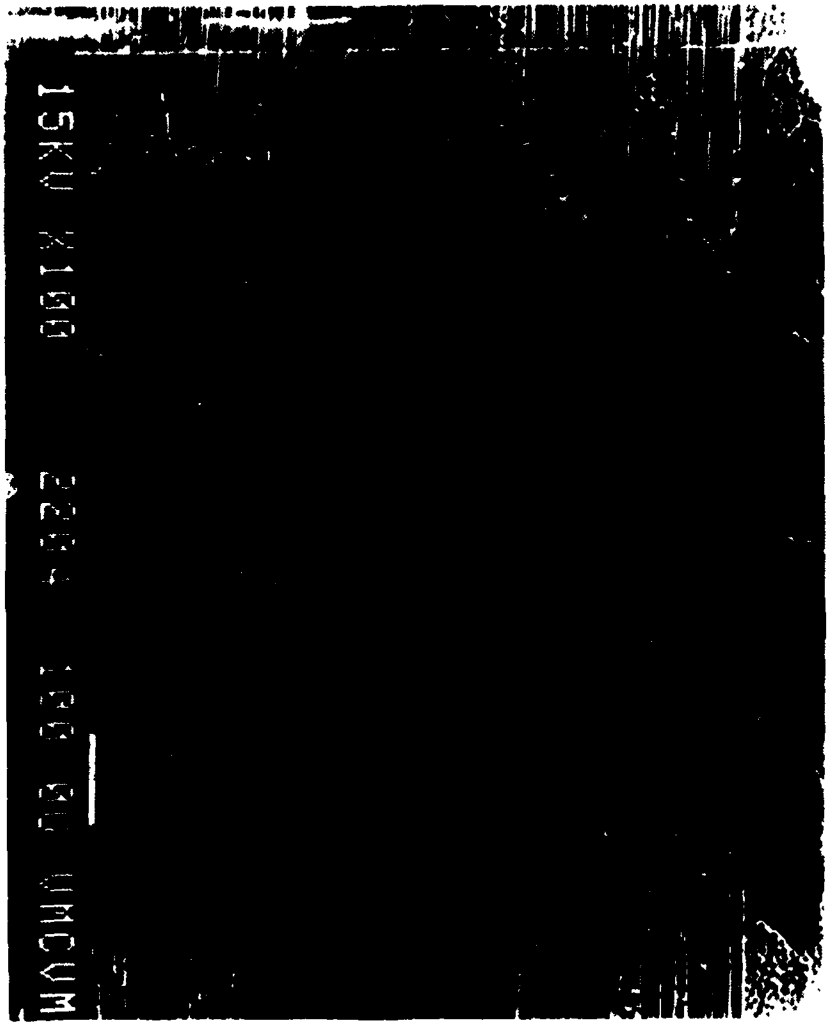

Figure 7a. Scanning electron micrograph of a replica of the edge of a $[0,\pm45,0]_{2s}$ laminate showing CDS cracks and some related fiber failures.

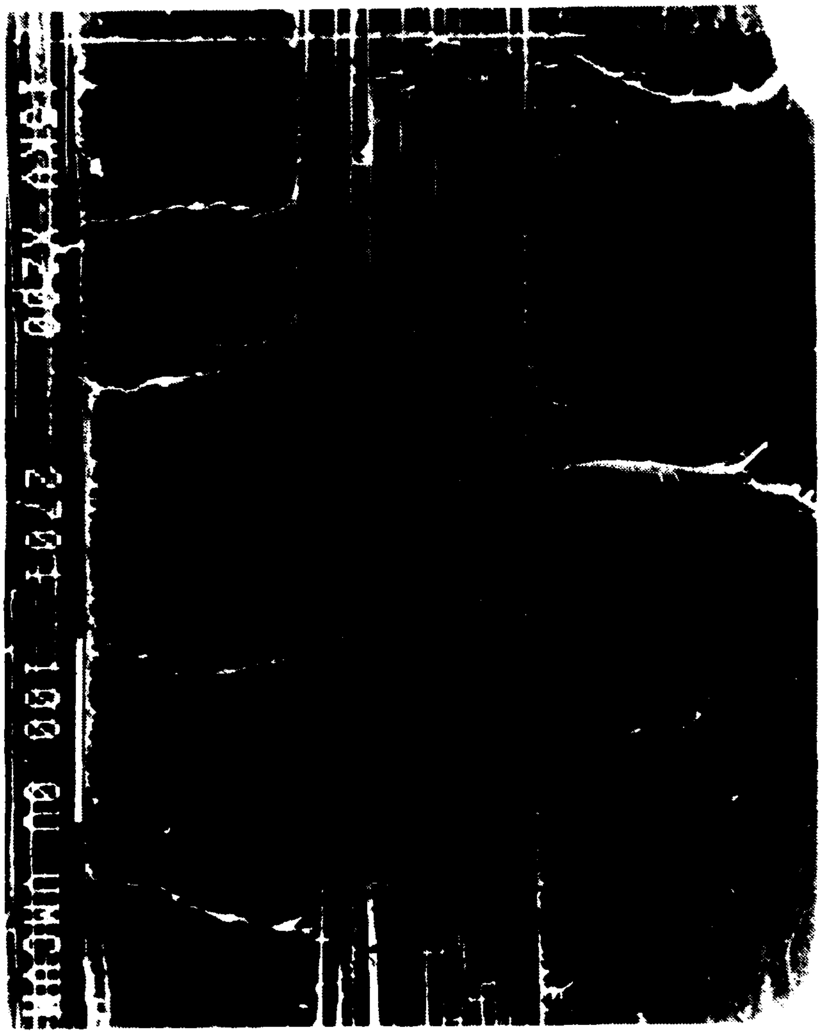

Figure 7b. Higher magnification of the pattern in Fig. 7a showing the association of fiber fractures with matrix cracks.

where the events occur. This was established by polishing the surface layer off and searching for fiber fractures beneath the thickness of a few fiber layers. The pattern of fiber fractures disappeared almost entirely when the surface layer was removed. It should also be mentioned that as the test continues, small local delaminations frequently develop at the tip of the matrix cracks and serve to isolate the localized pattern of fiber fractures from the influence of the stress state previously caused by the matrix cracks. In instances when edge delamination of a global nature occurs, that damage mode also serves to displace fiber fracture development as the dominant edge damage mode for the long life situation. Unfortunately, it is not possible, based on our investigation or other observations that the authors find in the literature, to state conclusively that the pattern of fiber fractures that develops at the edge of a specimen is or is not responsible for the initiation of the final fracture event.

While the problem of fiber fractures near the edge of a specimen is an important one, it is known that the edge problem is a rather specialized one and is not pertinent to many common practical engineering structures. The question remained, what is the fiber fracture situation in the interior of a laminate, and does it have any relationship to the clustered pattern that was observed at the edge. To investigate this situation, specimens which were cyclically loaded at levels which were about 60 to 70% of their static ultimate strength for periods of about 10^5 cycles (quite close to the final fracture event) were deplied and examined in the scanning electron microscope. In several instances, a solution of gold chloride and diethyl ether was applied to the specimens beforehand. That solution showed a surprising ability to penetrate the matrix cracks in the off-axis plies and, when the plies of the specimen were separated in the high temperature deply operation, gold from the gold chloride solution remained as a marker on the fibers of the zero degree plies in positions where matrix cracks were located in the adjacent off-axis plies. This technique was of great assistance to us in determining the relationship between the matrix cracks and the subsequent fiber fractures. Indeed, such a relationship was found. Figures 8 and 9 (taken from Ref. 9) indicate the nature of those results. The fiber breaks in the zero degree ply are seen to lie within a band of particles of gold marking the trace of an adjacent 90° transverse crack. In regions where no gold chloride is seen (as shown in Fig. 9) virtually no fiber fractures were found. This pattern was repeated for coupons taken from various points along the length and across the width of the specimen, and constituted a periodic array of zones of fiber breaks with a spacing equal to the characteristic damage state transverse crack spacing. Since the matrix cracks in the off-axis plies are regularly spaced in the CDS condition, the fiber fractures also occur in a periodic pattern of fiber breaks as shown schematically in Fig. 10. We have shown that the average spacing between the groups of fiber breaks is equal to the spacing between the matrix cracks in the ply that was adjacent to the zero degree ply under observation (Ref. 10). By removing several layers of zero degree fibers from the plies under observation, it was also determined that the fiber fractures were concentrated near the tip of matrix cracks in the adjacent ply in every direction, that is, the number of fiber fractures dropped off as a function of distance away from the matrix crack in the direction of the thickness of the adjacent zero degree ply as well as in

Figure 8. Scanning electron micrograph of a 0° ply taken from a fatigue damaged $[0,90_2]_s$ graphite epoxy laminate showing a fiber fracture zone.

Figure 9. Scanning electron micrograph of a 0 degree ply taken from a fatigue damaged $[0,90_2]_s$ graphite epoxy laminate showing a fiber fracture-free zone.

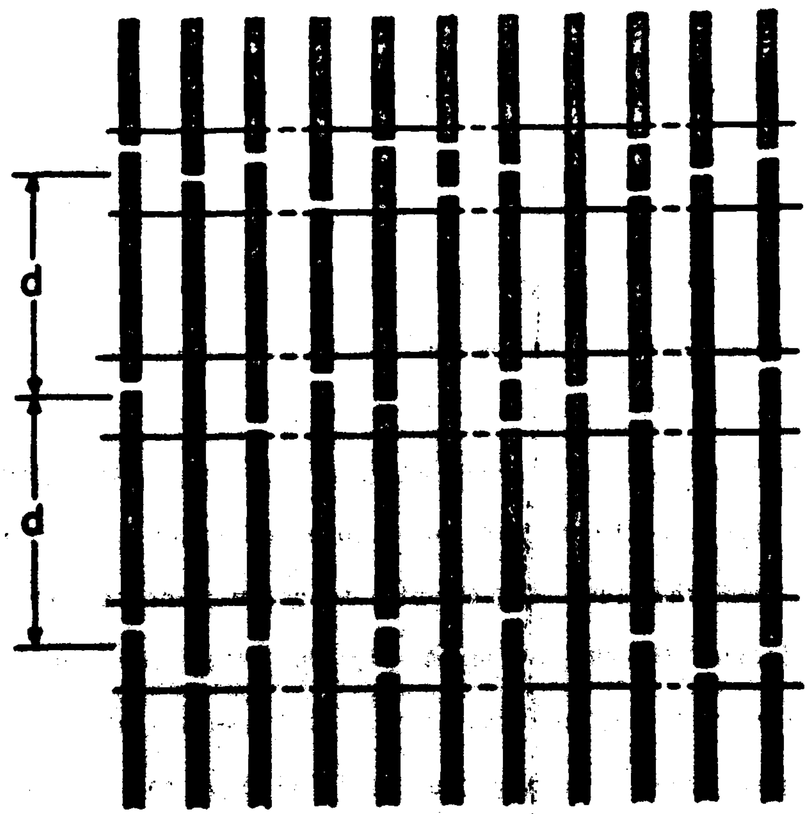

Figure 10. Schematic diagram of typical fiber break pattern.

the direction of the interface between the off-axis ply in which the matrix crack occurred and the zero degree ply in which the fiber fractures occurred (Refs. 4 and 10). The two salient features of these findings are the exclusive association of the fiber fractures with the matrix cracks in adjacent plies and the chronology of incidents of the fiber fractures. Since matrix cracking occurs early in the life of a long-term fatigue specimen or component, it would be expected that a preponderance of the fiber fractures associated with matrix cracking would also occur early in the test. This suspicion was verified by observing fiber fractures as a function of the number of applied cycles. An example of such data is reproduced in Fig. 11 from Ref. 10. In that figure, the number of fiber fractures in a $[0,90_2]_s$ laminate in a representative area are recorded for the first one-third, the second one-third, and the final third of testing. Also recorded is the number of fiber fractures that were observed after two fiber layers were removed from the zero degree ply interior surface, the surface previously adjacent to the off-axis ply in which the matrix cracks occured that caused the fiber fractures. It is seen that the number of fiber breaks does increase during the cyclic life of the material, but a preponderance of the breaks occur quite early, generally within the first third of the life. This finding was indeed a surprise since it was generally assumed (and has been widely stated in the literature) that fiber fracture was something that occurred only quite near to the end of the life of a specimen or component. Of course, it should be remembered that the number of fibers that are fractured in these observations is still a small fraction of the total number of fibers in a given zero degree ply. However, our results suggest that fracture of

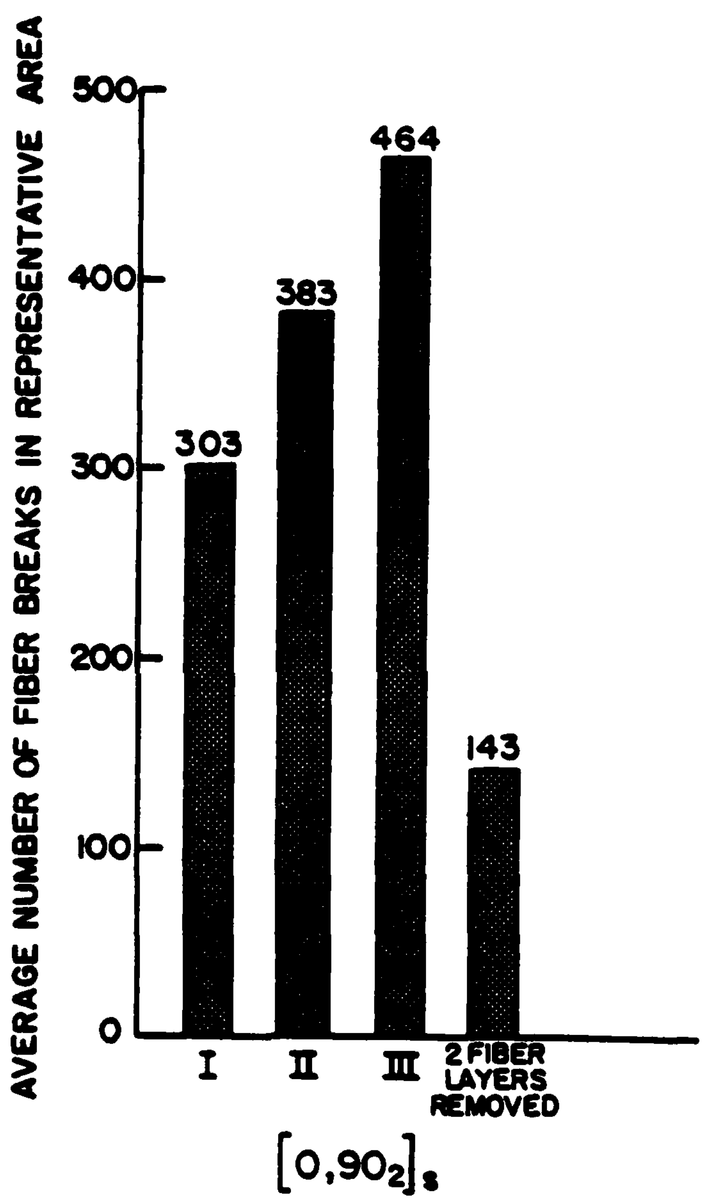

Figure 11. Total fiber breaks in a representative area of the zero degree ply taken from a fatigue damaged $[0,90_2]_s$ laminate at about 1/3, 2/3 and 1 lifetime of that laminate.

the total specimen occurs in the presence of only a small number of fiber fractures that occur within some critical region. Figure 12 presents a sample of our data that indicates the number of incidences of fracture of two fibers (doublets) or three fibers (triplets) or more fibers (multiplets) within a small (but arbitrary) distance from each other as a function of the number of cycles applied with similar information from the last third of testing for the zero degree ply with two fiber layers removed. Notwithstanding our rather arbitrary definition of singlets, doublets, etc., the number of groups of fiber breaks that we were able to find even in specimens which were severely damaged (for which the residual strength was reduced by as much as 20% or so) was surprisingly small, generally of the order of 4 or 5 or less.

As further evidence of the importance of matrix cracking to the fiber failure process, investigation of several other laminates with different stacking sequences produced substantially different numbers of fiber failures; the pattern of their occurrence was, however, essentially identical. While it is certainly clear that the number of fiber breaks is greatly influenced by the severity of the local stress concentration caused by matrix cracks, there is equally strong evidence that, during long-term fatigue loading, the number of fiber fractures is also closely related to the ability of the other (off-axis) plies to increase the compliance of the specimen by developing damage of one type or another to the extent that the specimen will fracture at the constant applied stress level. An equivalent statement would be to say that very few fiber fractures will occur during fatigue loading if the internal stress redistribution due to damage development in the off-axis plies is

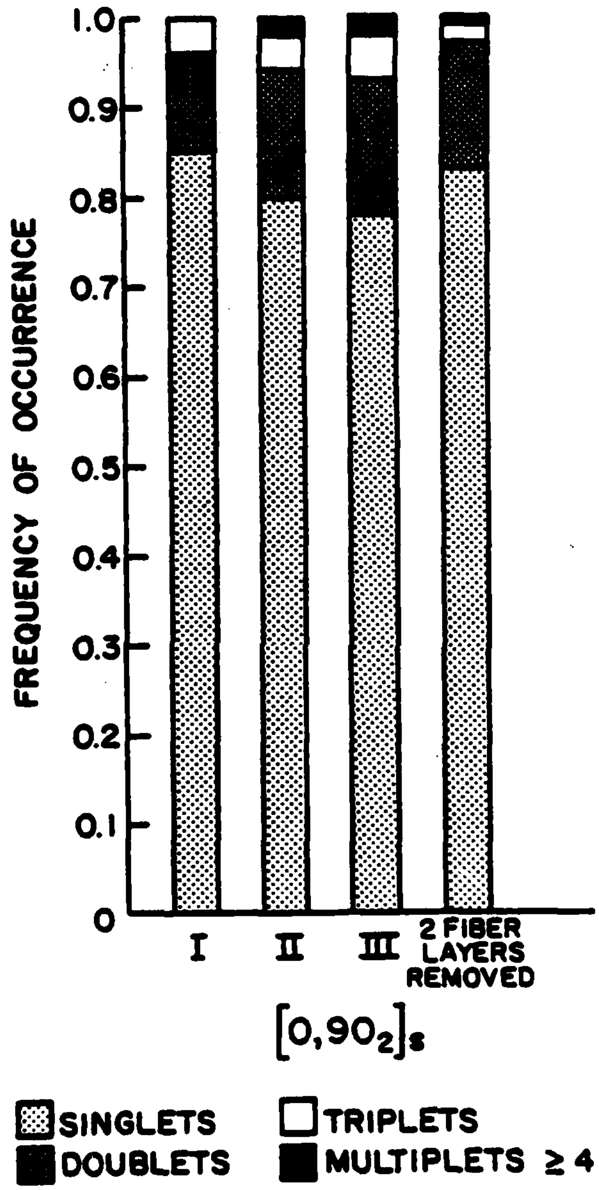

Figure 12. Distribution of "multiplets" in the specimen associated with the data shown in Fig. 10.

sufficient to increase the stress in the zero degree plies at the applied load on the laminate to a level which is sufficient to cause fracture of those zero degree plies and, consequently, fracture of the laminate. Still another essentially equivalent statement is to say that the number of fiber fractures that occurs under long-term fatigue loading in the zero degree plies is exactly equal to the number necessary to reduce the strength of the zero degree ply down to the current level of stress in that ply as determined from the applied laminate stress, the load sharing of the individual plies, and the stress redistribution caused by damage development in the off-axis plies and around the broken fibers (Ref. 10).

While a great number of unsettled issues remain regarding fiber fracture development, it is presently clear that these events are the beginning of the end. The fiber fractures that occur near matrix cracks in off-axis plies are the beginning of a localization pattern that eventually produces the fracture event in the laminate. Before the present investigation, such a statement would have been conjecture.

INTERFACE DAMAGE

One of the most fundamental problems with putting materials together to make composite materials is the inevitable tendency for them to come apart. In many cases, this tendency controls the long-term durability of composite components in service. Unfortunately, a preponderance of the prior investigations and the resulting literature deals only with the edge delamination problem, a very special case. In the opinion of the present investigators, the importance of the edge delamination problem is grossly exaggerated by coupon data. That problem has received a great deal of attention in the literature and was not addressed by the present investigation.

However, it was discovered during the course of the present investigation that delamination occurs in the interior of angly ply composite laminates under long-term cyclic loading in the final part of the life of those laminates in certain cases when strength reductions are large. The location of these internal delaminations is centered on positions where primary and secondary cracks cross. The primary driving force for the formation of these delaminations is thought to be the interlaminar normal stresses which are created at the interface between plies when matrix cracks terminate at that position. Those stresses have been estimated by Talug and Reifsnider using a three-dimensional finite difference scheme (Refs. 18 and 19). These regions of delamination have been observed in the $[0,90_2]_s$ graphite epoxy laminates. An example of such zone formation is shown in Fig. 13. Several zones are visible; one is indicated by an arrow. The occurrence of these delaminated areas would be little more than an interesting curiosity were it not for the remarkable capability of those regions to redistribute stress. Although the detailed three-dimensional stress state in the neighborhood of such delaminated areas has not been determined, a simple conceptual process will suffice to demonstrate the potential of that mechanism for stress redistribution. Table 1 records the stresses in each ply of a quasi-isotropic graphite epoxy laminate as determined from laminate analysis for several conditions, an undamaged condition, several conditions in which off-axis plies are cracked

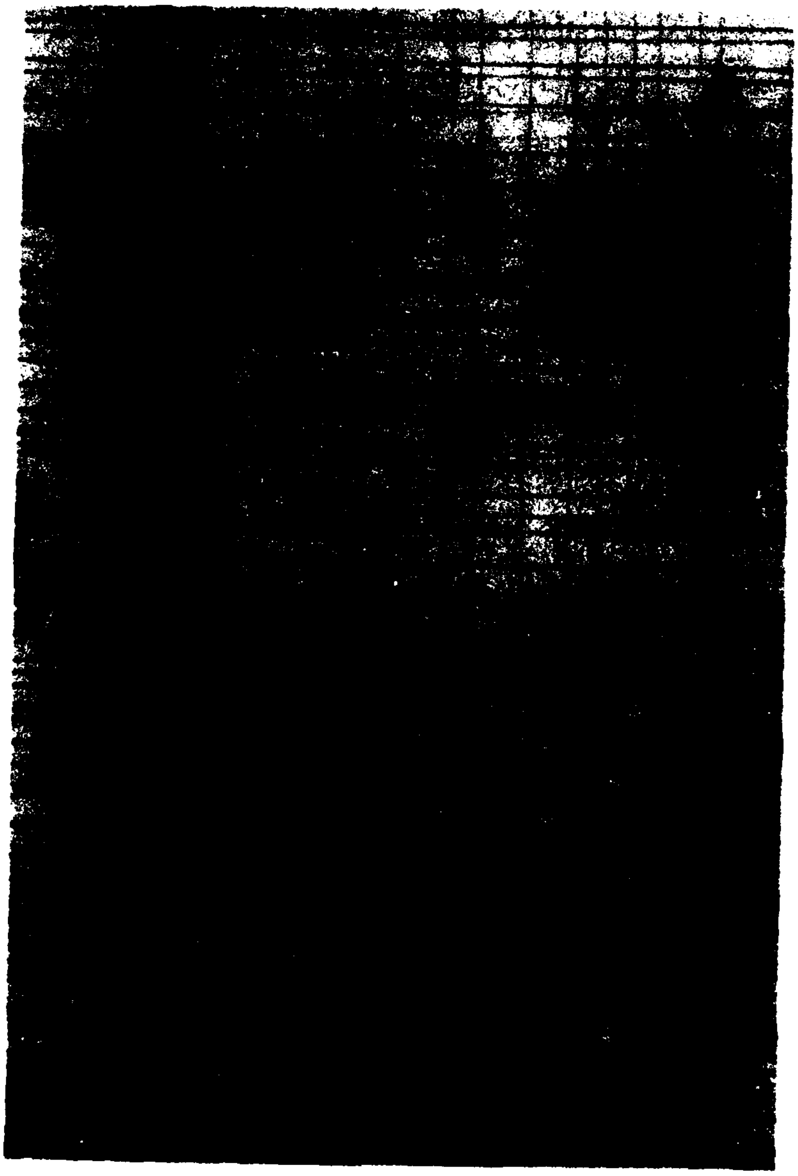

Figure 13. Radiograph showing several local internal delamination regions centered on crossed-crack positions (one shown by arrow).

simulated by reducing the in-plane shear stiffness and Young's modulus transverse to the fibers in those plies, and two situations where various parts of the laminate are delaminated. For illustration purposes, it is only necessary to notice the axial stress, σ_x, in the zero degree plies in each of the conditions described. That stress is 2632 for a 1,000 unit applied stress in the undamaged condition and increases to a value of 4,000 when all of the off-axis plies are delaminated and cracked. Even when a delamination occurs between the 90 and 45 degree plies in a quasi-isotropic laminate, there is a 19% increase in the axial normal stress in the zero degree ply. These results have a very profound consequence. Of all the individual and combined damage modes observed in the laminates investigated in this study, internal delamination is the only damage which has the capability to increase the average stress in the zero degree plies (and reduce the strength of the laminate proportionately) by magnitudes which are large enough to account for the large strength reductions of 20-50% observed in long-term cyclic loading of composite laminates. Hence, from the simple computations recorded in Table 1 and from a variety of additional more complex and precise computations (none of which are yet exact or otherwise completely accurate) it appears that internal delamination has the capability to reduce the strength of the laminate by the large amounts known to occur in long-term cyclic loading by stress redistribution alone, without any consideration of the zero degree ply degradation process due to such things as fiber fracture. This finding may, indeed, be the key to the mystery which motivated this entire investigation, the mystery of the fatigue damage mechanisms which cause large strength reductions under cyclic loading in angle ply high-modulus composite laminates.

Table 1. Stress distributions for matrix cracking and delamination

Laminate condition	Engineering moduli (GPa)				Ply stresses (MPa for 1000 MPa applied) 0			90			+45			−45		
	E_L	E_T	G_{LT}	ν_{LT}	σ_x	σ_y	τ_{xy}	σ_x	σ_y	τ_{xy}	σ_x	σ_y	τ_{xy}	σ_x	σ_y	τ_{xy}
Undamaged $[0, 90, \pm 45]_s$ laminate (T300-5208)	54.4	54.5	20.5	0.32	2632	−2.3	0	167	−797	0	600	400	418	600	400	−418
90° plies cracked	52.4	54.6	19.5	0.31	2735	0.1	0	0	−852	0	632	426	442	632	426	−442
90°, ±45 plies cracked	47.8	50.0	19.0	0.33	2993	−4.7	0	0	−1000	0	503	503	503	503	503	−503
Delaminated to [0, 90] and $[\pm 45]_s$ sublaminates (laminate stiffness $E_L \approx 46.1$)	77.2 15	77.2 15	4.1 37.2	0.04 0.81	3131	59	0	217	−59	0	325	0	136	325	0	−136
All off-axis plies delaminated and cracked	35.8	–	–	–	4000	0	0	0	0	0	0	0	0	0	0	0

DISCUSSION

A particular pattern of internal damage has been observed in several types of laminated high-modulus continuous (oriented) fiber composites subjected to cyclic loading which includes tensile load excursions at load (and strain) levels of the order of 40-60 percent of the ultimate strength of those laminates. That particular pattern is known to be associated with the strength, stiffness and life of such laminates and preliminary stress analysis suggests that the internal stress redistribution associated with such patterns is substantial, possibly large enough to account for the large strength reductions that are observed (Refs. 1-3). The distinctive nature and especially the three-dimensional aspects of that particular damage pattern were the principal object of this investigation. We have discussed the detailed nature of the specific modes of damage that make up such patterns. Our summary discussion will parallel the chronological development, i.e., we will discuss the individual damage modes that are observed in the sequence in which they are observed, and we will attempt to discuss the influence of each of those damage modes on subsequent events.

We are concerned with laminates which have plies that are oriented in various directions relative to a direction of tension and compression loading. If the amplitudes of the tensile load excursions are sufficiently high so as to produce any internal damage, matrix cracking will be observed. (This is true for notched or unnotched material and for metal matrix composites loaded above their fatigue limit as well as for polymer matrix materials in the unnotched condition which are the focus of our present attention (Refs. 7 and 8)). These matrix cracks are generally parallel to the fibers in the off-axis plies and terminate at the boundaries of those plies. For a given amplitude of stress-oscillation, the number of these cracks reaches an equilibrium value at a time which is fairly early in the life of the component, generally less than one-third of the total life. This saturation density of cracks occurs in a regular array in each cracked ply, called a characteristic damage state, which can be predicted by analytical schemes (Refs. 9-11). These cracks would also form in the same specimen under quasi-static loading to failure, and strength computations for that situation based on laminate analysis and a scheme for reducing the stiffness of cracked plies (the so-called discount method) suggest that the net-section stress redistribution in the neighborhood of such cracks controls the strength of the unbroken plies for that case. However, it has quite recently been found that during fatigue loading a certain number of fiber failures quickly develop near the tip of those cracks in the unbroken plies near the crack tips as suggested by the schematic diagram in Fig. 14 (Ref. 12). The number of such broken fibers drops off as a function of the distance into the thickness of the unbroken ply from the crack tip. While additional fiber failures continue to occur in these locations throughout the life of the material, the rate of fiber fracture seems to be very much greater in the early part of the life (up to about one-third of the total life span) than it is throughout the remainder of the loading history, even for load amplitude levels which are of the order of 50-60 percent of the static ultimate strength. Hence, the initiation of fiber failures in the neighborhood of matrix cracks can be thought of as the second chronological internal damage event. The total number of such fiber breaks varies greatly depending upon the laminate type and applied load level.

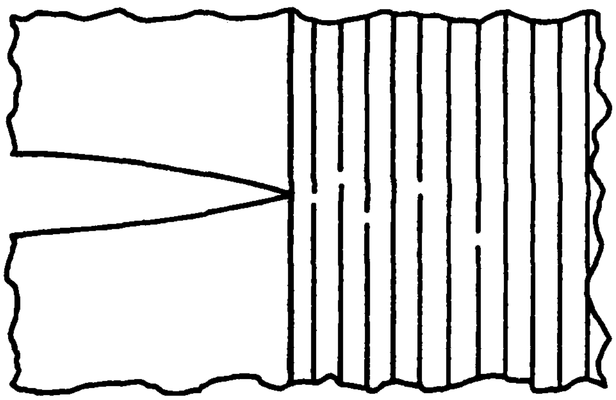

Figure 14. Schematic diagram of fiber breaks near the tip of a crack which terminates at a ply interface.

Each time a fiber fails, the matrix in the neighborhood of that fracture event is also damaged. It is difficult to sketch an idealized version of that damage. Figure 15 is an attempt to represent a cut-away view of a quarter-space section of local material near the end of a fiber failure. No attempt has been made to represent the microcracking and crazing of matrix material in that region, but debonding of the matrix from the broken fiber near the fracture position is indicated. While the presence of damage of these types in the regions around the fiber fractures has been firmly established over a period of many years, a complete and precise characterization of such damage under a variety of circumstances has not been made (Refs. 13-15). Nevertheless, it is clear that the fiber failures which occur in regions near matrix cracks induce matrix damage and debonding which is located preferentially near the interface between the cracked ply and uncracked ply.

In the situation under discussion (in the absence of edge-related damage propagation) a large part--generally more than half--of the fatigue life of the specimen or component following the early initiation period of the damage just discussed is spent in a relatively inactive period of damage development wherein additional fiber fractures and associated matrix damage occur at reduced rates near matrix crack tips, especially near ply interfaces. Generally speaking, as the final 25 percent or so of the life is approached, secondary matrix cracks begin to form in the ply adjacent to the initially cracked ply. In the instance when the global crack-opening normal stresses in that ply are tensile, these secondary cracks frequently grow rapidly through the thickness of the second ply and along the length of the specimen for a uniform stress case. In other instances these secondary cracks may be quite limited in extent and tightly grouped near the primary cracks in

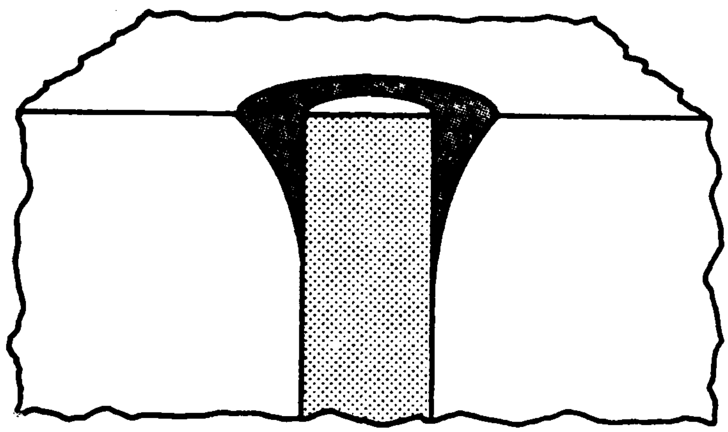

Figure 15. Schematic diagram of debonding at end of broken fiber.

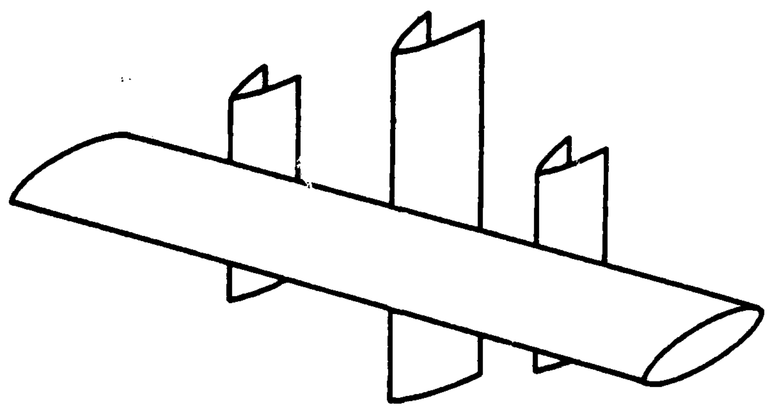

Figure 16. Schematic of crossed-crack problem in the interior of a $[0,90_2]_s$ laminate.

the first ply (Ref. 12). An idealized schematic of several such cracks appears in Fig. 16. The local stress state created at the intersection of these cracks is strongly three-dimensional and very difficult to determine. This problem is currently under study. However, it is clear that there is a strong tendency for the generation of interlaminar normal and shear stresses with significant magnitudes in that region (Refs. 1, 10 and 12). These stresses contribute to the development of damage in that area. In some cases, the interlaminar normal stresses cause local delamination as suggested by the schematic diagram in Fig. 17. This delaminated region may grow along the axes of the crossed cracks as suggested by the schematic in Fig. 18.

An example of this extreme case is shown in Fig. 19 which is an X-ray radiograph obtained from a $[0,90_2]$ symmetric graphite epoxy laminate specimen which had been cycled to about 80 percent of its total lifetime. The shaded regions at the intersection of longitudinal and transverse matrix cracks (the black lines in the vertical and horizontal directions) have been shown to be local internal delaminations of the type just described (Ref. 12, 14 and 1). One such region is indicated by an arrow. In laminates having other stacking sequences, the quasi-isotropic arrangement for example, the internal local delaminations have not been observed. In those instances, the scenario of localized damage development just described appears to terminate in the formation of secondary cross-cracks with the attendant local fiber failures and associated matrix damage in those regions.

The extent and degree of development of the various damage patterns discussed above is influenced by properties of the composite system, ply thickness dimensions, and cyclic load amplitude. However, the essential features of the pattern have been observed in graphite

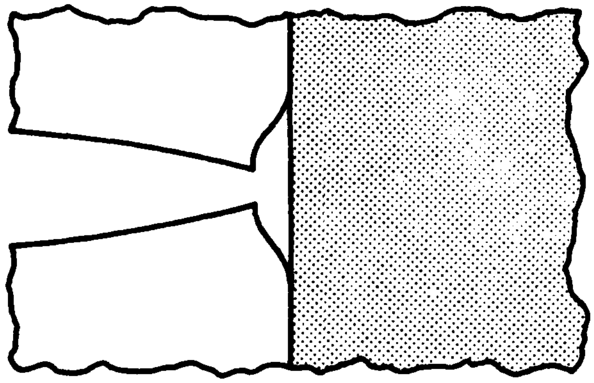

Figure 17. Local delamination at a ply interface near a matrix crack.

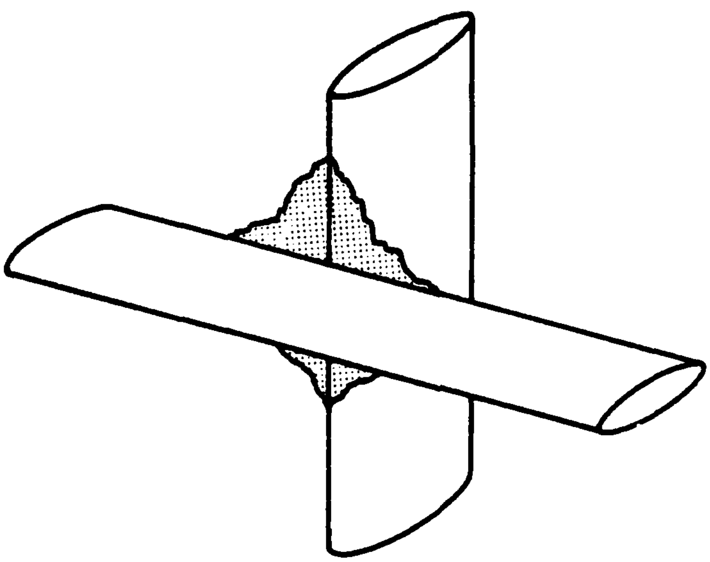

Figure 18. Localized matrix damage, fiber failure and delamination at the intersection of two matrix cracks at a ply interface.

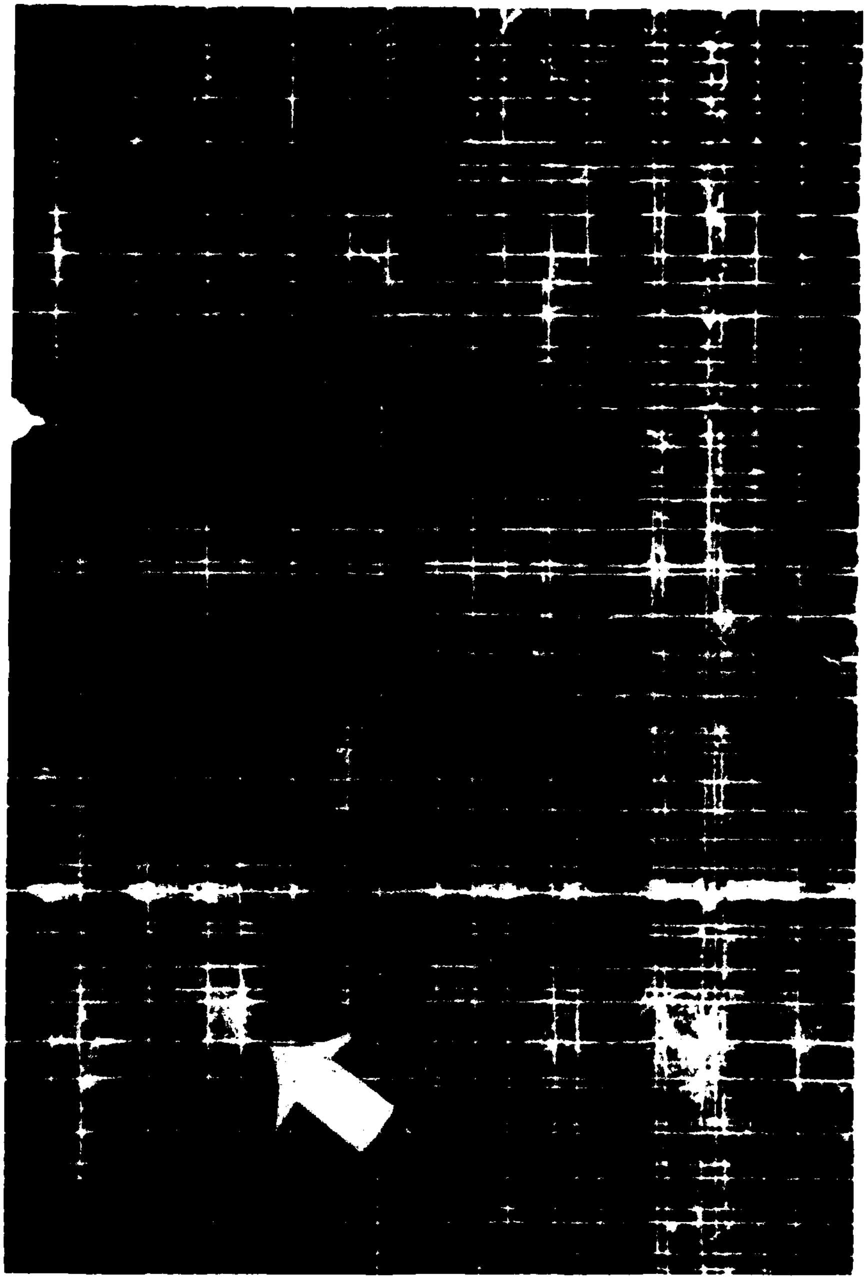

Figure 19. X-ray radiograph showing crossed-cracks in a $[0,90_2]_s$ graphite epoxy laminate with regions of local internal delamination near some of the crack intersections.

epoxy systems, glass epoxy systems, and certain metal matrix composites. These features have also been observed in widely different stacking sequences which induce different internal ply stresses at the global level. Based on our experience over a period of several years and the considerable evidence provided in the literature by other investigators, it appears that the internal damage pattern described above is a generic damage pattern for long-term cyclic loading of multiaxial laminates made from high-modulus continuous fiber materials. While investigations of these patterns are still in their early stages, it is known at this time that these patterns are widely distributed and are associated with changes in global stiffness, with local stress redistributions, and (by association) with fracture events (Refs. 1, 12 and 14). Developing an understanding of these patterns is especially challenging since they are distinctly three-dimensional and involve very complex internal geometric details. For example, in order to make an absolute association between these patterns and laminate residual strength, an accurate stress analysis of such a pattern must be obtained. Such an analysis is indeed complex. However, the nearly ubiquitous nature of this type of pattern in many situations involving the long-term cyclic loading of composite laminates suggests that further study of such patterns by investigators concerned with long-term strength and life is merited.

REFERENCES

1. K. L. Reifsnider, K. Schulte and J. C. Duke, "Long-Term Fatigue Behavior of Composite Materials," Long-Term Behavior of Composite Materials, ASTM STP 813 (1983).

2. K. L. Reifsnider, "Damage Mechanisms in Fatigue of Composite Materials," Proceedings, Third Risø Int'l. Symp. on Metallurgy and Materials Science, Fatigue and Creep of Composite Materials, Roskilde, Denmark, (1982) Risø National Laboratory, pp. 125-136.

3. K. Schulte, K. L. Reifsnider and W. W. Stinchcomb, "Entstehen Ausbreiton von Ermudungsschadigung in CFK," Proc. AVK Conf., Oct. 5-7, 1982, Freudenstadt, W. Germany.

4. R. D. Jamison, K. Schulte, K. L. Reifsnider and W. W. Stinchcomb, "Characterization and Analysis of Damage Mechanics in Fatigue of Graphite Epoxy Laminates," Proc. of Conf. on Damage Tolerance, Am. Soc. for Testing and Materials, 14 Dec. 1982, San Francisco.

5. A. L. Highsmith, W. W. Stinchcomb and K. L. Reifsnider, "The Effect of Fatigue-Induced Defects on the Residual Response of Composite Laminates," Proc. of Conf. on Damage Tolerance, Am. Soc. for Testing and Materials, 14 Dec. 1982, San Francisco.

6. A. L. Highsmith and K. L. Reifsnider, "Measurement of Nonuniform Microstrain in Composite Laminates," Composites Technology Review, 4, no. 1, (Spring 1982) p. 20.

7. K. L. Reifsnider and W. W. Stinchcomb, "Stress Redistribution in Composite Laminates," Composites Technology Review, 3, no. 1, (Spring 1981), p. 32.

8. R. D. Jamison, A. L. Highsmith and K. L. Reifsnider, "Strain Field Response of 0° Glass Epoxy Composites Under Tension," Composites Technology Review, 3, no. 4, (Winter 1981), p. 158.

9. K. L. Reifsnider and R. D. Jamison, "Fracture of Fatigue-Loaded Composite Laminates," Int'l J. of Fatigue, 4, no. 4 (1982) pp. 187-198.

10. R. D. Jamison, "Advanced Fatigue Damage Development in Graphite Epoxy Laminates," Ph.D. Dissertation, College of Engineering, Virginia Polytechnic Institute and State University, Blacksburg, VA, Aug. 1982.

11. K. L. Reifsnider, "Damage Mechanics and NDE of Composite Laminates", Mechanics of Composite Materials, Z. Hashin and C. T. Herakovich, eds., Pergamon Press (1982), pp. 399-420.

12. A. L. Highsmith and K. L. Reifsnider, "Stiffness-Reduction Mechanisms in Composite Laminates," Damage in Composite Materials, ASTM STP 775, K. L. Reifsnider, Ed., American Society for Testing and Materials (1982) pp. 40-62.

13. W. W. Stinchcomb and K. L. Reifsnider, "Fatigue Damage Mechanisms in Composite Materials: A Review," Fatigue Mechanisms, ASTM STP 675, J. T. Fong, Ed. (1979), pp. 762-787.

14. K. L. Reifsnider and R. D. Jamison, "Fracture of Fatigue-Loaded Composite Laminates," Int'l. J. of Fatigue, 4, no. 4, (Oct. 1982) pp. 187-197.

15. Damage in Composite Laminates, K. L. Reifsnider, ed., Special Technical Purlication 775, American Society of Testing and Materials (1980).

16. R. D. Kriz and W. W. Stinchcomb, "Effects of Moisture, Residual Thermal Curing Stresses, and Mechanical Load on the Damage Development in Quasi-Isotropic Laminates," STP 775, ASTM (1980) pp. 63-82.

17. J. E. Masters and K. L. Reifsnider, "An Investigation of Cumulative Damage Development in Quasi-Isotropic Graphite-Epoxy Laminates," STP 775, ASTM (1980) pp. 40-62.

18. K. L. Reifsnider, "Some Fundamental Aspects of the Fatigue and Fracture Response of Composite Materials," Proc. of 14th Annual Meeting of Society of Engineering Science, Lehigh University, Bethlehem, PA, Nov. 14-16, 1977.

19. A. Talug and K. L. Reifsnider, "Analysis of Fatigue Damage in Composite Laminates," Int'l. J. of Fatigue (Jan. 1980) pp.3-11.

20. K. L. Reifsnider, W. W. Stinchcomb and E. G. Henneke II, "Defect-Property Relationships in Composite Laminates", Report AFML-TR-76-81, Part IV, Air Force Materials Laboratory (Apr. 1979).

21. Damage in Composite Materials, ASTM STP 775, K. L. Reifsnider, Ed., American Society for Testing and Materials (1982).

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